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Division of Mechanical
Engineering Report

Rapport de la Division
de génie mécanique

1984/6

DM-4
NRC No. 23498

CARBON FIBRE REINFORCED ALUMINIUM MATRIX COMPOSITES

A CRITICAL REVIEW

M.U. Islam, W. Wallace

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LES MATÉRIAUX COMPOSITES À MATRICE D'ALUMINIUM RENFORCÉE DE FIBRES DE CARBONE
UNE CRITIQUE

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NRC No. 23498

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SUMMARY

In this report, the state-of-the-art technology on the carbon fibre reinforced aluminium matrix composites is reviewed. The available raw material, various fabrication techniques along with their effects on the properties of the composites produced and the manufacturing problems encountered are described in detail. General trends are outlined and further research and development work on composites containing pitch base carbon fibres in a matrix of high strength aluminium alloy is recommended.

SOMMAIRE

Ce rapport présente une critique de la technologie de pointe des matériaux composites à matrice d'aluminium renforcée de fibres de carbone. Les matières premières disponibles, les techniques et les problèmes de fabrication et leurs effets sur les propriétés des composites sont examinés en détail. Les recherches en cours et les développements futures sont soulignés. En conclusion, le rapport recommande de poursuivre un programme de recherche et développement sur les composites à matrice d'alliages d'aluminium à haute résistance renforcée de fibres de carbone produites à partir de brai.

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CARBON FIBRE REINFORCED ALUMINIUM MATRIX COMPOSITES
A CRITICAL REVIEW

1.0 INTRODUCTION

Metal matrix composites offer unique properties that make them extremely attractive for a number of industrial and aerospace applications. They are distinguished from the more developed resin matrix composites by virtue of their metallic nature in terms of physical and mechanical properties and their ability to lend themselves to conventional metal working processes. In most cases the mechanical properties of metal matrix composites far exceed those of the matrix itself. Properties such as very high specific stiffness and near zero thermal expansion are common in both metal matrix and resin matrix composites, however, the definite advantages of metal matrix composites include:

1. Excellent thermal conductivity, good electrical conductivity and non-flammability.

2. Minimal moisture absorption.

3. Minimal outgassing and no contamination by the products of outgassing.

4. Minimal contamination by organic fluids such as fuels, lubricants, solvents and cleaning agents.

5. High temperature capability.

6. High matrix shear strength, ductility and abrasion resistance.

7. Ability to be coated, joined, formed and heat treated by conventional metallurgical processes.

Reinforcing fibres are available in a variety of materials, including carbon, silicon di-oxide, boron, silicon carbide, S-glass, beryllium, steel and tungsten. Most of these fibres are coated with some other material to form a diffusion barrier, in order to avoid degradation of the fibre by fibre/matrix interaction, and/or to enhance fibre/matrix compatability. Similarly, a number of metals can be used as matrix material. It is evident that the properties of metal matrix composites may be tailored to meet a specific need through the proper selection of the reinforcing fibres, the volume fraction of the reinforcing fibres and the matrix material. In Figure 1, the specific mechanical properties of the available metal matrix systems are compared with some of the resin matrix carbon systems and some alloys.

Metal matrix composites are broadly classified into two categories:

1. Fibre reinforced composites. This class is further subdivided into continous or discontinuous fibres with uni-directional or bi-directional reinforcement.

2. Particle or whisker reinforced composites. Here the subdivision is based on random or preferred orentation of the whiskers or particles.

However in this investigation, only aluminium matrix composites with continuous reinforcing fibres of carbon will be reviewed. Particular attention will be given to the available fibres, alloy matrices, fibre surface coatings, manufacturing processes and their influence on the properties of the resulting carbon/aluminium composites.

2.0 THE FIBRES

There are various high performance continuous fibres commercially available. In Table 1, the fibres which can be used with an aluminium matrix are listed along with their mechanical properties. It should be noted that these properties are those for the fibres only. In a real structure a matrix is present and the fibres may not be aligned uni-directionally. Hence the properties of the resulting composites will be inferior to those of the fibres by an amount depending on the volume fraction of the fibres, the distribution and precise geometrical configuration of the fibres in the composite, the strength of the fibre/matrix interface and the nature of the matrix.

As can be seen from Table 1, various fibres have different densities. In most practical applications it is desirable to minimize weight. Hence an important feature is their specific mechanical properties, i.e. strength/density and modulus/density. Figure 1 shows the specific strength and specific modulus values for the fibres usually used for reinforcing aluminium. In Table 1 the axial thermal expansion characteristics of various fibres are also listed. It should be noted that carbon fibres have a negative axial thermal expansion coefficient. It is this feature that makes it possible to design and build composite structures with near-zero thermal expansion characteristics by appropriate fibre lay-up in the structure.

2.1 Carbon Fibres

Carbon fibres are generally produced by the thermal decomposition of various organic fibres. Rayon, polyacrylonitrile (PAN) pitch, polyesters, polyamides, polyvinyl alcohol, polyvinylidene chloride and phenolic resins have all been considered as potential precursor material for producing carbon fibres[2]. Of these, rayon, polyacrylonitrile (PAN) and pitch have been found to offer the greatest potential as starting material. The process to convert precursor material into fibre is different for each starting material. However, all the conversion processes have a common final step of graphitization of the fibres[1].

The final step of graphitization in the manufacture of graphite fibres has a significant effect on the properties of the fibres. It can be seen from Figures 2 to 5, that the modulus of both PAN and pitch base fibres increases with increasing graphitization temperature. However, the strength of PAN base fibres shows a peak at around 1200°C, while the strength of pitch base fibre increases with heat treatment temperature. The higher degree of graphitization also tends to increase the negative thermal expansion coefficient and electrical conductivity[2]. The graphitic nature of the fibres can result in the formation of an electro-chemical couple when carbon fibres are in contact with metals. In the presence of an electrolyte this may lead to a rapid corrosion unless appropriate precautions are taken. The properties of some selected carbon fibres available in USA are listed in Table 2 along with the name of their manufacturer.

Rayon base carbon fibres have been widely used with aluminium matrix[5,6,7,8,12,13,19,21,26,30,34], and have provided the most consistent behavior in terms of dispersion within the aluminium matrix and in terms of composite properties. However, rayon base fibre is very expensive and the production is being discontinued[2,26], it is for this reason that rayon base carbon fibres are not considered as viable reinforcing agents for aluminium.

PAN base fibres are further subdivided on the basis of their mechanical properties into the following three general types:

a) TYPE 1 — High modulus fibres.

b) TYPE 2 — High strength fibres.

c) TYPE 3 — Lower cost variants.

Tensile strength and modulus of PAN base fibres depend on the final graphitization temperature (Figs. 2 and 3). The reactivity of PAN base carbon fibres is somewhat unpredictable, when incorporated in aluminium alloys. For instance, PAN base carbon fibre T-300 reacts extensively with aluminium alloys resulting in fibre degradation and lower strength composites, while HM-3000, which is also a PAN base fibre, yields some very high strength composites. This difference in behavior may be related to the different surface characteristics of these fibres. T-300 is a relatively low modulus fibre (227 GPa), which results from the relatively low final heat treatment temperature. T-300, therefore, is not highly graphitized and the surface is chemically reactive. HM-3000, on the other hand, is a high modulus fibre indicating a relatively high final heat treatment temperature, a high degree of graphitization, and a much less reactive surface. PAN base carbon fibres GY-70, T-300, HM-3000 (HMS type with about 3000 fibres per tow) and Celion HT (Table 2) have all been used as reinforcing agents with various aluminium alloys[5,7,8,9,12,21,22,23,24,25].

Pitch base carbon fibre tows (or bundles) are the cheapest and most easy to infiltrate with molten aluminium, and the composite wires produced from them possess sound structure with a uniform fibre dispersion[5,8,12,21]. However, strength values for the composite wires and plates are quite low because of the current low strength of the fibres. This problem is being addressed and higher strength fibres may be available in the future[2,5,26].

Carbon fibres are available in various forms: continuous, chopped, woven fabric or mat. Tows, yarns, rovings, and tapes are the most common forms of carbon fibre available. Depending on the organic precursor and manufacturer, typical filament count in a tow may vary from 400 to 10,000 or as high as 160,000. Generally the price increases as the fibre count decreases. The ultra-high modulus PAN base fibre GY-70, manufactured by Celanese Corporation USA, contains about 384 fibres per tow. For use in metal matrix composite, aluminium and magnesium alloy infiltrated, coated, carbon fibre tow 'wire', woven cloth, tape and braid preforms are marketed by Materials Concepts USA[1]. These can be directly used for hardware fabrication by a variety of consolidation processes described in Section 4.3.

2.2 Boron Fibres

Boron fibres are manufactured by a continuous chemical vapour deposition process using an electrically heated substrate core. Boron fibres are available with tungsten wire or carbon filament cores with diameters in the range of 100-200 μm.

Boron filaments react with titanium and aluminium at an unacceptably rapid rate[2], at the temperatures used for the manufacture of composites with these metals. The reactions are temperature dependant and they severely degrade the mechanical properties of the fibres. Thus it is extremely important to have a diffusion barrier layer on the boron filament, so as to improve the filament properties in titanium matrix composites, and to permit the use of high temperature in the manufacturing of aluminium matrix composites.

Two diffusion barrier layers are commercially available for use with boron filaments, silicon carbide (Borsic) and boron carbide (B_4C). This latter coating enhances the strength of the fibres rather than reducing it as with Borsic[2]. Other barrier coatings including boron nitride (BN), tungsten carbide (WC) tantalum carbide (TaC), hafnium carbide (HfC) and titanium carbide (TiC) are also in the developmental stage. Some of these have been used to coat SiC filaments for use in superalloy matrices[2,3].

The mechanical properties and thermal expansion coefficients of boron filaments are listed in Table 1.

2.3 Silicon Carbide Fibres

Chemical vapour deposited, continuous silicon carbide fibres are available either with tungsten or carbon cores. These filaments are intended primarily for metal matrix applications, and

have good elevated temperature property retention, good resistance to oxidizing atmospheres and they react slowly when brought in contact with molten metals. Their mechanical properties and thermal expansion coefficients are listed in Table 1.

2.4 FP Alumina

FP alumina is a continuous polycrystalline α-alumina fibre, available in diameters of about 20 μm, and is finding use where its electrically insulating and high temperature properties can be exploited. Other fibres of polycrystalline β-alumina and filaments consisting of ultrafine crystallites of spinel structure are also being developed[4].

3.0 THE MATRIX

The matrix material has the dual function of maintaining the component shape and the transference of load to the fibres via shear processes at the interface. This function is critically affected by the strength of fibre/matrix bond, which will be discussed in detail later in this review for carbon/aluminium systems. Aluminium, because of its low density, availability and low cost is extremely attractive for use as a matrix material.

The most common aluminium alloys used with carbon fibres are: 201[5,6,7,8,12,13,21,22,23,25,26,30], 6061[7,8,9,10,14,16,17,18,21,26,30,34] and 1100[8,10,16,18,19,26,30]. Pure aluminium[11,28], 356[9,14,26], 413[19,26], 5056[26], 5154[26], A1-5Mg[24], A1-13Si[34], A1-10Mg[34] and A1-12Si[27] have also been used as matrices. The properties of these aluminium alloys and some other alloys which have not been used as matrices are listed in Table 3. It can be seen that 7000 series aluminium alloys, which have the highest tensile strength in the heat treated condition, do not appear to have been used as matrices for carbon fibre reinforced composite systems.

4.0 FABRICATION METHODS

The objective of a fabrication process is to combine the fibres and the matrix in a useful form, such that:

a) The fibres are introduced in the matrix without mechanical damage.

b) The fibres are aligned and distributed uniformally.

c) Adequate bonding between fibre and matrix is obtained.

Some fibres are very sensitive to any kind of mechanical damage that may be introduced either by simple mutual contact of the fibres or rubbing, scratching or tool handling. As their strength is dependent on their surface integrity, they must be handled with great care in order to retain their characteristics.

The properties of a fibre composite depend on the uniformity of distribution and the degree of fibre alignment. Fibre distribution and alignment is usually obtained either by hand lay-up of alternate layers of fibres and matrix foils, or by using a modified lathe to wind fibres on matrix foil mounted on a mandrel.

The most important and the most difficult problem associated with the fabrication of composite materials is the achievement of good bonding between fibre and the matrix. This should be obtained without any detrimental chemical attack or reaction at their interfaces and is essential to achieve good load transfer between matrix and fibre. Without bonding, there is no transfer of load to the fibre and hence no reinforcement. Good bonding may be promoted by[20]:

a) Mutual interfacial diffusion restricted to the formation of a solid solution.

(b) A superficial non-detrimental chemical reaction.

(c) Perfect interfacial contact between the components.

These favourable bonding parameters can be promoted by various primary manufacturing processes for metal matrix composites.

The primary manufacturing methods for metal matrix composites can be generally classified as liquid, deposition or diffusion processes (Table 4). In practice however, a combination of these processes is used. For instance when the matrix is formed by either deposition (vapour or electroplating) or a liquid — phase process, the composite is usually densified by a diffusion process.

Deposition processes in conjunction with liquid metal infiltration and diffusion processes have been widely used for the production of carbon fibre reinforced aluminium matrix composites, and are fairly well established. The major disadvantage of using liquid state processes for the fabrication of carbon/aluminium composites is the requirement of higher processing temperatures, which promote excessive fibre/matrix interaction. Solid state methods such as a deposition process in conjunction with a diffusion process, although not as well established, have the potential of producing carbon/aluminium composites with better mechanical properties, since the interface reaction can be controlled more easily at lower temperatures. However, one other advantage of liquid metal infiltration processing is that theoretically fibres can be infiltrated by any alloy composition.

4.1 Manufacturing Problems

The major problems encountered in the fabrication of carbon/aluminium composites are:

a) Reaction between aluminium and carbon.

b) Poor wetting of carbon fibres by molten aluminium.

c) Oxidation of the carbon fibres.

In the following sections these problems will be discussed in detail.

4.1.1 Interface Reactions

It is well established that at temperatures higher than 500°C, carbon reacts with aluminium to form aluminium carbide (Al_4C_3). This interface reaction is diffusion controlled and can degrade the mechanical properties so severely as to render the material useless for practical applications[30].

Shorshorov et al[27] studied the interface reaction in carbon/aluminium composites fabricated by vacuum-compression infiltration of a carbon tape (type of fibres not indicated) by Al-12% Si alloy. They reported that increasing infiltration pressure, time and temperature increases the quantity of Al_4C_3 at the interface, and temperature has the most significant effect.

Motoki et al[28] reported that carbon fibres (high strength type, precursor not mentioned) and vacuum evaporated pure aluminium react to form aluminium carbide after vacuum annealing for four hours at 550°C or higher and for one hour at 620°C or higher. The ultimate tensile strength of pure aluminium evaporated carbon fibres after vacuum annealing for 9 hours at 600°C and 16 hours at 630°C, was reduced to 61% and 29% respectively of that of the original carbon fibres before annealing. The UTS decrease reported for ion plated 5056 aluminium alloy on similar fibres was 66% and 51% for the specimens annealed at 600°C for 9 hours and at 630°C for 16 hours respectively. This lower reduction of UTS for 5056 alloy as compared to pure aluminium was attributed to alloying elements present in 5056 alloy. However, it was concluded that aluminium carbide formation is a time and temperature dependant process, and that aluminium carbide also degrades the properties of the fibre by roughening the fibre surface, which causes stress concentrations.

Baker et al[31,32] also reported that high strength carbon fibres coated by pure aluminium using a vacuum vapour deposition technique, exhibited lower room temperature strengths after exposure to temperatures ⩾475°C for 100 hours in vacuum. The extent of degradation in strength increased with an increase in annealing temperature up to 650°C, and a minimum UTS value of about 30% of the original strength value was produced. Aluminium coated high modulus carbon fibres also showed a similar behavior, with a reduction in UTS after anneals at ⩾550°C, and a minimum UTS value at 45% of the original strength. The degradation is attributed to the formation of aluminium carbide at the interface as no such behaviour was observed with uncoated fibres with similar high temperature exposure. However, high modulus fibres were shown to have a greater resistance to degradation by aluminium carbide formation as compared to high strength fibres. The UTS of aluminium coated fibres was also shown to depend on the inverse square root of the aluminium coating thickness.

In a study of the carbon/aluminium interface reactions, Khan[30] prepared diffusion couples by vacuum deposition of aluminium on to carbon in various crystallographic forms. It was reported that the chemical reaction begins at about 500°C in vacuum and the reaction rate increases with increasing temperature. The reaction product was also identified as Al_4C_3, with a hexagonal structure (a_o = 3.32 Å, c_o = 24.89 Å). The reaction zone thickness at the interface was shown to vary linearly with the square root of the reaction time, and increasing temperature increased the rate of the reaction (Fig. 7). This further confirms that the reaction process is dfffusion controlled, and no appreciable effect of various crystallographic forms of carbon on the interface structure was observed.

Pepper et al[34] exposed rayon base T-50 fibres with coatings of aluminium alloys A13, 220 and 6061, to high temperatures for five minutes, and reported no degradation in strength below 680°C.

Asanuma et al[29], while investigating the use of roll diffusion bonding process for carbon/aluminium composites, reported that degradation of high strength PAN base fibres in air begins at 450°C, and at 600°C the strength was decreased by 50% in just one hour. Oxidation of the fibres was reported to be the cause.

4.1.2 Wetting

It is well known that carbon fibres are not spontaneously "wet" by molten aluminium, and this is because the contact angle between carbon and aluminium is greater than 90 degrees[15,33]. Without wetting, infiltration is not possible[20], and therefore wetting must be improved by:

a) Modification of the chemical composition of the surface of the fibres.

b) Modification of the chemical composition of the liquid matrix.

c) Increasing the working temperature.

d) Modification of the working atmosphere.

However, for carbon/aluminium systems the wetting characteristics of the fibres are usually improved by the modification of the surface of the fibres by various coatings (Section 4.2). These coatings not only improve the wettability of the carbon fibres, but it is also believed that they retard the detrimental chemical reaction between aluminium and carbon at the typical fabrication temperatures.

4.1.3 Oxidation of the Carbon Fibres

Carbon fibres are extremely resistant to oxidation (even in excess of 2000°C) in inert environments, but they readily oxidize at moderate temperatures in air. Thus for high temperature composite fabrication processes, the carbon fibres must be protected from oxidation.

Galasso and Pinto[38] and Baker et al[39] investigated the oxidation resistance of carbon fibres and aluminium coated carbon fibres when exposed at elevated temperatures in air. It was found that the rate of reduction of fibre cross-section (high modulus fibres) was much less in coated than in uncoated fibres at temperatures above 500°C. Figure 6 shows the percentage carbon weight loss plotted against time at 475°C and 525°C for coated and uncoated high modulus carbon fibres.

However, Baker et al[39] also reported that the composites made from high modulus carbon fibres and aluminium could be exposed in air at 400°C for one week without any loss of strength.

4.2 Production of the Carbon/Aluminium Precursor Wire

In the production of carbon/aluminium composites, the first stage is usually the production of precursor wire, which is then consolidated to form simple shapes in the second stage. To overcome the problems listed in Section 4.1, the carbon fibres are coated and depending on the coating, are then either passed through molten aluminium to produce carbon/aluminium precursor wire or used directly as precursor wire for subsequent consolidation.

The most common coating for the production of carbon fibre precursor wire is of titanium di-boride[5,6,7,8,9,12,14,30,33,34,35]. The process involves a continuous chemical vapour deposition (CVD) operation, that deposits a thin layer of titanium and boron onto multi-filament carbon fibre tows by the reduction of titanium tetra-chloride ($TiCl_4$) and boron tri-chloride (BCl_3) with metallic zinc vapours. This thin coating, less than 200 Å thick, activates the surface of the carbon fibres to permit wetting and minimizes fibre degradation. The coated fibres are then passed through molten aluminium to produce carbon/aluminium precursor wire.

Another process known as the soldium process, exploits the good wetting characteristic of carbon fibre by molten sodium and has also been used to prepare carbon/aluminium precursor wire[8,19]. In this process carbon fibres are passed through molten baths of sodium at around 550°C, molten tin with small amount of magnesium at about 600°C and finally through molten aluminium at 20-50°C above liquidus temperature, for 1-10 minutes per bath[19]. Sodium forms a thin surface layer on the carbon fibre, which acts as a wettable and protective coating. Immersion in molten tin dissolves and displaces sodium, leaving a composite that when solidified, consists of a nearly pure tin matrix with a thin layer of sodium rich metal remaining at the fibre/matrix interface. Final immersion in molten aluminium alloy dissolves and displaces the tin, since aluminium and tin are also completely miscible, leaving carbon fibres imbedded in a matrix of the aluminium alloy.

Other processes for producing carbon/aluminium precursor wire (without titanium di-boride coating) include:

a) Electroplating of nickel on the carbon fibres and then passing through molten aluminium[8].

b) Vapour depositing of nickel on the carbon fibres[5].

c) Ion plating (IVD) of carbon fibres with aluminium[24,37].

d) Plasma spraying of carbon fibres with powdered aluminium[29].

The properties of the carbon/aluminium precursor wire produced by various methods will be discussed later in this review.

4.3 Consolidation of the Composite Laminate

In this final stage of primary manufacturing of carbon/aluminium composites, the precursor wire is consolidated into simple shpaes such as sheets, bars or rods. Continuous processes, including hot roll bonding[11,29] and hot drawing[9,14] and batch type process of hot pressure bonding[6,7,8,10,13,18,23,24,26,30,34,35,36] in vacuum or in air, have all been used as primary consolidation processes.

The most commonly used process for the consolidation of carbon/aluminium precursor wire is hot pressure bonding in vacuum. In this process, the precursor wire is pre-collimated either by hand or by winding on a modified lathe, to provide desired spacing. Mono-layered or multi-layered sheets or plates are then obtained by high temperature and high pressure consolidation of alternate layers of matrix foil and pre-collimated carbon/aluminium precursor wire.

Various temperatures have been used for the hot pressure bonding of carbon/aluminium precursor wire. Pepper et al[34] reported that pressing above the liquidus temperature resulted in segregation of the fibres in the composite, a large amount of shrinkage porosity and in some cases degradation of the carbon fibres.

Satisfactory composites have been obtained by partial liquid phase hot pressing[34,36], in which the precursor wire is bonded under pressure at temperatures above the solidus temperature in the two phase solid/liquid state in vacuum. Increasing consolidation temperature increases the fibre volume fraction in the composite due to the expulsion of the molten matrix alloy. This strength increasing effect is however, negated by the chemical interaction between aluminium and carbon, which is also temperature dependant. The washing effect during the expulsion of increased amount of liquid at higher temperatures also tends to aggravate the chemical interaction. For example, 1600 ppm of Al_4C_3 were detected in 6061/Thornel 50 system, when pressing was done at 630°C, and only 625 ppm of Al_4C_3 was obtained at 615°C. The fibre volume fractions obtained at 630°C and 615°C were 45% and 31% respectively, but the tensile strengths obtained were similar in both cases. The fibre fraction was found to be independant of pressure from 3.5-35 MPa[36].

Pressing at lower temperature below solidus requires the use of higher pressures, which may cause fragmentation of fragile carbon fibres and a significant loss of composite properties[34]. However, Khan[30] and others[8,13,23,24] reported that satisfactory properties can be obtained in composites produced by solid state diffusion process in vacuum.

Other processes that have been used for the consolidation of carbon/aluminium precursor wire are the roll bonding and hot drawing. The properties of the composites produced by these techniques will be discussed later in this review.

5.0 MECHANICAL PROPERTIES AND THE EFFECTS OF FABRICATION PROCESSES

Generally, the rule of mixtures can be applied to predict the mechanical properties of carbon/aluminium composites. The rule of mixtures (ROM) is expressed as the contribution to the strength and modulus of the composite by the two components based on their relative volumes. This rule, as commonly applied, assumes that there is no synergistic or interaction effect between the components and is expressed as:

$$\sigma_c = v_f \sigma_f + (1 - v_f) \sigma_m$$

where:

σ_c = strength of the composite,

σ_f = strength of the filament,

σ_m = strength of the matrix, and

v_f = volume fraction of the filaments.

Most of the tension test data on composites has been presented in light of this rule of mixtures, and the composites are rated in terms of the percentage of the rule of mixtures strength that they have achieved.

In the following sections, the properties obtained using different techniques of precursor wire production and subsequent consolidation will be reviewed.

5.1 Precursor Wire

The mechanical properties of the carbon/aluminium precursor wire produced by various methods are listed in Table 5. It can be seen that most of the data available is on composites produced by using rayon base, T50 carbon fibres with a number of aluminium alloys. As mentioned earlier, rayon base fibres are the most expensive fibres with cost almost 10 times that of PAN base fibres, and for this reason they are being taken off from the market[2,26]. Thus, rayon based fibres cannot be considered as viable reinforcing agents, although consistent mechanical properties are obtained in composites produced with these fibres. Table 5 also show that for the production of carbon/aluminium precursor wire, the titanium-boron vapour deposition/liquid metal infiltration method is commonly employed. Various fibre cleaning methods and pyrolytic carbon coatings have also been tried prior to titanium-boron deposition. Some comparative data is also available for the precursor wire produced by the sodium process, the nickel coating/liquid metal infiltration process and vacuum vacuum vapour deposition (IVD) of pure aluminium powder on the fibres.

For T50/201 system, Ti-B vapour deposit/liquid metal infiltration method consistantly provided strength over 80% of the rule of mixtures prediction. Pyrolytic carbon coating prior to Ti-B vapour deposition, increased the strength of T50/201 composite by 10%. This improvement might have resulted from either the improved fibre fracture strength, caused by the healing of surface flaws, or from the additional layer of oriented carbon, which limits fibre/matrix interaction[8]. The sodium process yielded the lowest strength composites (66% of the rule of mixtures prediction), and excessive fibre/matrix interaction, probably due the absence of a diffusion barrier, was reported to be the cause[8].

T-300 fibre with 201 aluminium showed a marked effect when a pyrolytic carbon coating was applied prior to the Ti-B vapour deposit step. The strength increased from 19% to 92% of the rule of mixtures strength[8]. However, with other matrices such as 356 and 6061, T-300 fibre yielded strengths up to 108-119% of the rule of mixtures predicted value when a Ti-B vapour deposited coating was used in conjunction with liquid metal infiltration process[9].

Sullivan and Raymond[33] studied the effects of the liquid metal infiltration process on the surface of T-50 (Rayon base), Modmor Type-1 (PAN base), Hercules AS (PAN base) and Thornel Type P (Pitch base) carbon fibres. The fibres after infiltration were extracted by leaching with dilute hydochloric acid, and the surfaces of these fibres and as received fibres were then compared by using carbon replica electron microscopy techniques. It was observed that T-50 and Thornel Type P fibres were minimally attacked by liquid metal infiltration, while both types of PAN base fibres showed a significant loss of surface features coupled with a massive amount of Al_4C_3 formation. It was however concluded that carbon fibres which have seen higher graphitization temperature produce composites with good mechanical properties and low Al_4C_3 contents. Moreover, the mechanism of fibre degradation during the liquid metal infiltration process is high temperature oxidation coupled with attack by molten aluminium.

Baker et al[31] showed that 98-134% rule of mixtures strength could be obtained by vacuum vapour deposition of pure aluminium powder on to high modulus, high strength PAN base fibre. Nickel coating and liquid metal infiltration of high modulus and high strength PAN base fibres provided only 34-35% rule of mixtures strength[8,11].

Only two studies are reported on pitch base fibres[5,8]. Both the studies used Ti-B vapour deposit/liquid metal infiltration method and over 61% rule of mixtures strength was reported with 201 aluminium alloy matrix.

5.2 Consolidated Composites

In Table 6 the tensile properties of the consolidated composites produced by hot pressing are listed along with the consolidation parameters. It can be seen that consolidation temperatures both in the solid state and in the two phase (solid/liquid) regions have been utilized. Maximum tensile strength of 1014 MPa was obtained in HM 3000/201 composite (fibres coated with TiB_2), when the pressing was done at 568°C for 25 minutes at 24 MPa pressure. This corresponds to a value of 98% rule of mixtures strength[8].

Generally an average of about 58% rule of mixtures strength is attained in consolidated composites as compared to an average of 82% in the precursor wire. This decrease in strength accompanying the consolidation of the precursor wire is usually attributed to the increased carbon/aluminium reaction at the interface, producing increased amount of Al_4C_3, although the amount of Al_4C_3 in the precursor wire is not generally measured. However, Watts[26] reported that vacuum hot pressing of titanium di-boride coated fibres either above or below the solidus temperature increases the amount of Al_4C_3 to a great extent. The composite wire usually has 150-1500 ppm of Al_4C_3, while in consolidated composites 176-4150 ppm of Al_4C_3 is usually found (Table 5). This is also consistent with the fact that the formation of Al_4C_3 is a diffusion controlled reaction and the temperatures at which consolidating is done increase the reaction rate. Moreover it can be concluded that the Ti-B coating on the carbon fibre is not a very effective barrier coating at the composite consolidating temperatures.

Gigerenzer et al[9] used hot drawing to consolidate T300/6061 and T300/356 aluminium precursor wire having a Ti-B chemical vapour deposited coating. The processing temperature used was 504 + 12°C, which is below the solidus temperature for both 6061 and 356 aluminium alloys. Consolidation was achieved at a drawing rate of around 5 cm/s^{-1}, and the best results were obtained at 5-15% reduction with 35-40 volume percent fibres. Ultimate tensile strengths between 717 and 924 MPa corresponding to 75-86% rule of mixture values and tensile moduli in the range of 98.6 to 140 GPa were reported. Metallographic examination of the composites revealed many areas containing discontinuous fibre sigements without any voids, indicating that flow of the matrix had occurred around the broken fibres. It was concluded that consolidation of carbon/aluminium precursor wire by hot drawing results in high translation of fibre properties due to minimum mechanical damage induced during processing.

Asanuma and Okura[29] used roll diffusion bonding in air to consolidate layers of carbon/aluminium composite tape, produced by plasma spraying of aluminium on carbon fibres (spreaded tows). Although strengths equivalent to 76% of the rule of mixtures value were obtained in composite specimen containing 12 volume percent fibres, excessive fibre breakage during rolling was reported. Composite oxidation, non-uniform fibre distribution and fibre/fibre contact were also reported to be the factors contributing to lower mechanical properties of the composites produced.

The transverse strengths of carbon/aluminium composites are listed in Table 7, along with their fabrication method. It can be seen that the transverse strengths range from 9.7 to 55.8 MPa for all the composites, which is very low. The highest transverse strength of 55.8 MPa was observed in T300/201 composite, but the longitudinal strength reported for the same composite was only 212 MPa corresponding to 19% of the rule of mixtures strength. An Al_4C_3 content of 3280 ppm was reported[8].

Amateau et al[8] studied the effect of processing on the transverse strength of carbon/aluminium composites produced by various methods. It was observed that the transverse strength is related to the degree of reaction between fibre and the matrix. The fibres which reacted most extensively with the matrix had the greatest transverse strength as shown in Figure 8, in which the transverse strength is plotted against the Al_4C_3 content. Fabrication and thermal processes that promote fibre/matrix interaction also increase the transverse strength, but do so at the expense of longitudinal strength. Matrix composition was found to have very little effect on the transverse strength and the composites with low volume fraction of fibres showed highest transverse strengths.

In an attempt to increase the transverse strength, surface and interlaminar foils of various compatible aluminium alloys have been placed between the layers of the precursor wire for consolidation[26]. These foils provide a continuous metallic cross-section possessing a transverse strength equal to the matrix aluminium alloy. Higher transverse strengths were obtained in the composites, however, the use of these foils reduces the longitudinal strengths as well as the transverse and longitudinal moduli[26].

Harrigan[35] studied the chemistry of the interface in T50/6061 composite after various thermal and mechanical treatments by using an ion microprobe mass analyzer (IMMA). The interface concentration of chromium, silicon, magnesium, titanium and boron was studied in the following conditions:

1. As received composite precursor wire.
2. Precursor wire treated for 1000 hours at 465°C.
3. Precursor wire creep tested at 450°C at 400 MPa for 50 hours.
4. Composite bar consolidated at 615°C, 0.15 MPa for 5 minutes.
5. Composite bar consolidated at 615°C, 0.25 MPa for 10 minutes.

The results obtained are summarized in Figure 9. It can be seen that heating, straining and consolidating the wire cause changes in the character of the interface. The changes in the concentration of chromium, silicon and titanium at the interface were reported to have a minimal effect on the strength of the composite. However, the increase in magnesium content at the fibre interface above five atomic percent and decrease in boron content at the fibre surface below 20 atomic percent were associated with low longitudinal strength.

Marcus et al[6], while analyzing the fracture surface of T50/201 composite by scanning auger microscopy (SAM) also observed a similar increase in magnesium content at the interface. The magnesium at the interface was reported to be in totally oxidized form, while aluminium was present in both oxide as well as in metallic form. It was concluded that the reaction zone of the carbon/aluminium composite consisted of an oxide rich region adjacent to the fibre with a titanium and boron-rich region between the oxide and the matrix as illustrated in Figure 10. The transverse fracture path was found to meander between the fibre, the reaction zone and the matrix (Fig. 9). However, the highest transverse strength was associated with those composites for which the fracture path was predominantly through the oxide layer rather than between fibre/oxide interface.

6.0 CORROSION

Carbon/aluminium composites are inherently prone to galvanic corrosion, since carbon fibres provide a good cathode and have a free corrosion potential about one volt higher than aluminium. Metzger and Fishman[40] reviewed the corrosion of aluminium matrix composites recently and concluded that corrosion in T50/201, T50/201-1100 (201 aluminium in wires and 1100 in interlayer foils) and T50/6061 composites proceeded preferentially at foil-foil, wire-foil and wire-wire interfaces. The process is initiated by a pit penetrating the surface foil or at the end of a plate. Once initiated, the corrosion spreads rapidly, and exfoliation and disintegration of the composite can result from the wedging action of the corrosion product formed in the crevices.

The corrosion of clad and coated carbon/aluminium composites was also studied by various investigators (as mentioned in Ref. 40). Composites with 6061 or 2024 claddings were exposed to alternate immersion in seawater or marine exposure. The behavior was similar to that of the bulk alloy, the claddings were not breached. Paints (polyurethane, chlorinated rubber and epoxy) were effective in protecting against corrosion. Electroplated nickel also proved to be quite successful, but once the nickel coating was breached, the matrix becomes a sacrificial anode and corrosion was rapid.

7.0 SUMMARY, CONCLUSIONS AND RECOMMENDATIONS

The mechanical and physical properties of carbon fibres make them very attractive for use as reinforcing fibres in various metallic as well as non-metallic matrices. Based on the precursor material, carbon fibres can be classified as Rayon base, PAN base or Pitch base fibres. PAN base fibres can be further subdivided as:

a) TYPE-1 High modulus fibres.

b) TYPE-2 High strength fibres.

c) TYPE-3 Lower cost variants.

For carbon fibre reinforced aluminium matrix composites, most of the effort so far has been to use Rayon base fibres. Rayon base fibres, due to their comparatively low reactivity with aluminium, have provided composites with consistent properties, but unfortunately these fibres are very expensive and are being taken off from the market.

PAN base carbon fibres react somewhat unpredictably when incorporated in aluminium matrix. Tensile strengths ranging from 19 to 134% of theoretical values have been reported in the aluminium matrix composites with PAN base fibres (Table 5). The reactivity of these fibres is believed to be controlled by the final graphitization temperature used during their production. The higher the graphitization temperature, the less reactive will be the fibres with aluminium. Thus higher (rule of mixture) strengths are reported in composites with TYPE-1 or high modulus fibres, which have seen higher graphitization temperatures.

Although very little work has been done on pitch base carbon fibre/aluminium composites due to low available fibre strength in the past, they appear most promising as potential reinforcing fibres. Pitch base fibres are the cheapest and high strength and high modulus fibres are now commercially available. As can be seen in Figures 2 to 5, strength and modulus for pitch base fibres increase with final graphitization temperature, while for PAN base fibres the strength reaches a peak at about 1200°C and then start to decrease, however modulus continues to increase with temperature. Thus, theoretically the production of even higher strength and higher modulus pitch base fibres is possible and there are indications in the literature that the fibre manufacturers are also working in this direction. Even at the present time, commercially available pitch base fibre P-100 produced by Union Carbide (Table 2) has the highest modulus of 689 GPa as compared to 517 GPa of PAN base high modulus fibre GY 70. The strength of pitch base fibres is also comparable to PAN base fibres. The other advantage of high graphitization temperature may be the reduced reactivity of these fibres with aluminium. However, a lot more work is required to characterize the behavior of pitch base fibres when incorporated in aluminium matrix.

For carbon/aluminium composites, aluminium alloys 201 and 6061 are most commonly selected as the matrix material. Other alloys such as 1100, 220, A13 and 356 have also been used as matrices. However no investigations employing high strength 7000 series alloys have been reported, although one of the major reasons for incorporating carbon fiabres in an aluminium matrix is to improve the strength. As can be seen in Table 3, 7075 has substantially higher tensile strength than any other aluminium alloy used as a matrix. According to the rule of mixtures, the higher the strength of the starting constituents, the higher will be the strength of the resulting composite. In Figure 11 the predicted strength of pitch base P-100 fibres (data about four years old, higher strength pitch base fibres amy have been developed by now) in 7075 aluminium matrix is plotted as a function of fibre volume fraction. At 0.35 fibre volume fraction, the strength predicted by the rule of mixtures is 1096 MPa, and is comparable to the strengths obtained in other systems.

The production of carbon/aluminium composite requires the use of high temperature and pressure. Besides physical difficulties in obtaining proper fibre distribution and alignment due to lateral movement of fibres during consolidation, the major chemical problems encountered in the production are:

a) Interface reaction between carbon and aluminium.

b) Lack of wetting.

c) Oxidation of the fibres at processing temperature.

The major efforts to overcome these difficulties have been in the direction of developing coatings for carbon fibres. The most commonly applied coating is of TiB_2 by chemical vapour deposition. Nickel coating by electro-plating and Na-Sn-Mg coating by sequential immersion in molten baths has also been employed.

In Figure 12, a simplified flow chart for the production of carbon/aluminium composites is shown. Fibres which are coated with nickel, plasma sprayed by powdered aluminium or ion vapour deposited by aluminium are usually consolidated with matrix foils, while the fibres coated with TiB_2 or Na-Sn-Mg are passed through molten aluminium and are consolidated with or without matrix foils.

The mechanical properties of the precursor wire as well as consolidated composites vary widely (Table 5 and 6), and fabrication technique appears to have a major influence. Processes that expose fibres to excessively high temperatures, promote increased carbon/aluminium reaction and hence lower strengths. Since interface reaction is a diffusion controlled reaction, proper time and temperature is required to control the thickness of the reaction zone. Lower temperature deposition and diffusion processes although not yet established, have future potential of producing composites with improved mechanical properties.

To summarize, as illustrated in this review, most of the data is on composites prepared with rayon base T-50 fibres which are being taken off from the market. Thus further work is needed to evaluate the behavior of Pitch and PAN base fibres when incorporated in a high strength aluminium alloy matrix, and properties established before these composites become of significant value as engineering materials.

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9.0 ACKNOWLEDGEMENT

The authors would like to thank Mr. G. McGregor, General Manager, Manufacturing Technology Centre, for proof reading and suggesting some necessary changes in the final draft of this report.

TABLE 1

TABLE SHOWING VARIOUS PROPERTIES OF THE AVAILABLE FIBRES FOR USE IN ALUMINIUM MATRIX COMPOSITES

Fibre	Precursor	Tensile Strength GPa	Tensile Modulus GPa	Density g/cm^3	Thermal Exp. Coef. ppm/K
Carbon	PAN	1.7-3.2	170-517	1.7-2.0	-0.5
	Rayon	0.6-3.2	41-525	1.6-1.8	to
	Pitch	1.1-2.1	379-689	2	-1.3
Silicon	W Core	3.1	420	3.15	4
	C Core	3.45	400	3.0	—
Boron	W Core	3.52	400	2.49	5
	C Core	3.32	380	2.25	—
Borsic	W Core	2.9	400	2.5	4.9
Alumina	—	1.38	380	3.9	6.8

TABLE 2

PROPERTIES OF CARBON FIBRES AVAILABLE IN USA

Manufacturer	Fibre Name	Precursor	Tensile Strength GPa	Tensile Modulus GPa	Density g/cm^3
Great Lakes	Fortafil 3	PAN	2.48	207	1.71
Corp. USA	Fortafil 5	PAN	2.76	331	1.80
Celanese	Celion GY-70	PAN	1.86	517	1.96
Corp. USA	Celion 6000	PAN	2.76	234	1.76
	Celion 3000	PAN	2.76	234	1.76
	Celion 1000	PAN	2.48	234	1.76
	Celion 12K, 3K	PAN	3.24	234	1.77
Hercules	AS	PAN	3.10	220	1.77
Incorp.	HTS	PAN	2.76	248	1.80
USA	HMS	PAN	2.34	344	1.86
Union	Thornel 50	Rayon	2.20	393	1.67
Carbide	Thornel 300	PAN	2.65	227	1.75
USA	Thornel 75	Rayon	2.65	524	1.82
	Thornel B	PAN	3.2	290	1.71
	P 55	Pitch	2.07	379	—
	P 75	Pitch	2.07	517	—
	P 100	Pitch	2.07	689	—

TABLE 3

PROPERTIES OF SOME ALUMINIUM ALLOYS

Aluminium Alloy	Density gm/cm^3	Modulus GPa	Tensile Strength MPa	Thermal Expansion Coefficient $\times 10^{-6}$ 20-100°C
Pure A1	2.71	69.0	82.8	13.2
1100	2.71	69.0	89.7	13.1
2024	2.77	73.1	186.3 (annealed) 448.5 (T3)	12.9
5052	2.69	70.3	193.2 (annealed) 262.2 (H34)	13.2
5056	2.63	71.07	289 (annealed) 414 (H38)	13.4
6061	2.71	69	124.2 (annealed) 310.5 (T6)	13.0
7075	2.80	71.7	227.7 (annealed) 572.7 (T6)	13.1
201	2.80	—	448.5 276 (T6)	10.7

TABLE 4

GENERAL PROCESSES USED FOR THE PREPARATION OF FIBRE-REINFORCED METAL MATRIX COMPOSITES

Liquid Processes	Deposition Processes	Diffusion Processes
Infiltration of liquid maxtrix metal between fibres. (Usually followed by pressure at temperature above the solidus.) Pressing of coated firbres and matrix (foils or powder) at a temperature above the solidus of the matrix.	Electrodepositions of matrix around pre-arranged fibres. (Usually followed by hot pressing either above or below the solidus temperature.) Plasma spraying of matrix around pre-arranged fibres (generally followed by pressing). Vacuum vapor deposition of matrix around fibres (followed by pressing).	Pressing of bare or coated fibres between foils of matrix metal (below solidus).

TABLE 5

MECHANICAL PROPERTIES OF CARBON/ALUMINIUM PRECURSOR WIRE PRODUCED BY DIFFERENT METHODS

Precursor Wire	Production Method	Volume % Fibres %	Tensile Strength MPa	Tensile Strength GPa	% ROM Strength %	Reference
T50/201	Ti-B/LMI	32	655-690	138		5
T50/201	Ti-B/LMI	—	743	—	84	8
T50/201**	Ti-B/LMI	27.0	565	160		36
T50/201	Ti-B/LMI	27-32	655-828		86-97	26
T50/201	Na process/LMI	—	407	—	66	8
T50/201**	Ti-B/LMI pre-cleaned fibres	27.3	635	161		36
T50/201	Pyrolytic carbon coating on fibre then Ti-B/LMI	—	920	—	94	8
T50/201	Fibres hydrogen pre-cleaned then TI-B/LMI	—	763	—	82	8
T50/1100	Ti-B/LMI	21-23	621-759		98-100	26
T50/1100	Na process/LMI	13.3	317		96.9	19
T50/A413	Ti-B/LMI	26-33	483-710		65-81	26
T50/413	Na process/LMI	28	731		105	19
T50/5056	Ti-B/LMI	30	593		69	26
T50/5154	Ti-B/LMI	28.32	621-759		85-89	26
T50/6061	Ti-B/LMI	27-34	586-793		77-88	26
T50/6061**	Ti-B/LMI	29.5	655			35
T50/6061**	Ti-B/LMI	27.5	415	144		36
T50/6061*	not indicated	3.3	643			34
T50/6061**	Ti-B/LMI pre-cleaned fibres	28.9	578	157		36
T50/356	Ti-B/LMI	28-33	483-690		63-79	26
T50/A13*	not indicated	21.2	693			34
T50/A13*	not indicated	33.0	612			34
T50/A13*	not indicated	33.6	616			34
T50/A13*	not indicated	25.1	523			34
T50/A13*	not indicated	31.0	670			34
T50/A13*	not indicated	29.4	661			34
T50/220*	not indicated	35.1	569			34
T50/220*	not indicated	7.3	520			34
T50/220*	not indicated	34.5	484			34
HM3000/ 1100	Sodium process then LMI	9.3	210		87.7	19
GY70/201	Ti-B/LMI	37-38	793	207		25
GY70/201	Ti-B/LMI	35.40	931-965	220		5

TABLE 5 (Cont'd)

MECHANICAL PROPERTIES OF CARBON/ALUMINIUM PRECURSOR WIRE PRODUCED BY DIFFERENT METHODS

Precursor Wire	Production Method	Volume % Fibres %	Tensile Strength MPa	Tensile Strength GPa	% ROM Strength %	Reference
T300/1100	Na process/LMI	11.3	98		31.8	19
T300/201	Ti-B/LMI	—	212	—	19	8
T300/201	Pyrolytic carbon coating on fibre then Ti-B/LMI	—	1131	—	92	8
T300/356	Ti-B/LMI	35-40	1103-1310	96-138	112-119	9
T300/6061	Ti-B/LMI	35-40	1034-1276	110-138	108-119	9
HTS/Pure Al	Vacuum vapor dep. 46-540 nm		2080-2890	196-243		31
HMS/Pure Al	Vacuum vapor dep. 46-540 nm		1650-1800	278-290		31
HM/201	Ti-B/LMI	35-40	931-1000	165		5
VS054/201	Ti-B/LMI	48-52	1035	345		25
HM1000/201	Nickel coating then LMI	—	138	—	34	8
HT11/A14Ni	Nickel coating then LMI	20-45	200-450	—	35	11
Pitch/201	Ti-B/LMI	35-40	469-524	138		5
Pitch/201	Ti-B/LMI		313		61	8

LMI — Liquid metal infiltration.

ROM — Rule of mixtures.

* — Prepared by Materials Concepts Incorp., USA.

** — Prepared by DWA Composite Specialities Incorp., USA.

TABLE 6

CONSOLIDATING PARAMETERS AND THE MECHANICAL PROPERTIES OF THE CONSOLIDATED CARBON/ALUMINIUM COMPOSITES

Consolidation Parameters	Tensile Strength MPa	Fibre Content %	Al_4C_3 Content ppm	% ROM Strength %	Reference
T50/356 (Ti-B)					
626 C, 2.76 MPa, 10 min	296-407	37-57	800-3000	30-31	26
T50/A413 (Ti-B)					
598 C, 4.14 MPa, 10 min	269-365				26
T50/6061 (Ti-B)					
575 C, 21 MPa, 40 min	434	27			30
615 C, 4.14 MPa, 10 min	490-676	28-33	300-1000	63-77	26
615 C, 0.15 MPa, 5-10 min	586	30	1145		35
615 C, 0.25 MPa, 5-10 min	449	32	772		35
615 C	558	31	625		36
621 C, 20.7 MPa, 30 min	379-517	26-32	250-500	52-61	26
626 C, 4.14 MPa, 10 min	559-724	41-50	750-2500	53-59	26
630 C	560	45	1600		36
670 C, 3.4-6.9 Mpa	631	42.5			34
675 C, 3.4-6.9 MPa	440	26.7			34
685 C, 3.4-6.9 MPa	517	30.0			34
690 C, 3.4-6.9 MPa	291	26.0			34
T50/201 (Ti-B)					
555 C, 21.0 MPa, 40 min	531	27			30
560 C, 20.7 MPa, 30 min	517-586	28-32	400-800	66-69	26
570 C, 24.0 MPa, 25 min	743	32	1012	84	8
604 C, 4.14 MPa, 10 min	552-690	29-34	340-700	68-77	26
T50/5056 (Ti-B)					
621 C, 20.7 MPa, 30 min	331	22.3	1050-1450	50	26
T50/1100 (Ti-B)					
587 C, 20.7 MPa, 40 min	276	35	4150	30	26
643 C, 13.8 MPa, 5 min	303	42	2015	28.5	26
T50/5154 (Ti-B)					
598 C, 4.14 MPa, 10 min	524	27	250	76	26
615 C, 4.14 MPa, 10 min	524-600	37-40	600-750	55-58	26
T50/A13					
645 C, 3.4-6.9 MPa	396-676	36-46			34
650 C, 3.4-6.9 MPa	584	43.9			34

TABLE 6 (Cont'd)

CONSOLIDATING PARAMETERS AND THE MECHANICAL PROPERTIES OF THE CONSOLIDATED CARBON/ALUMINIUM COMPOSITES

Consolidation Parameters	Tensile Strength MPa	Fibre Content %	Al_4C_3 Content ppm	% ROM Strength %	Reference
T50/220					
645 C, 3.4-6.9 MPa	573	26.6			34
650 C, 3.4-6.9 MPa	588-677	27-37			34
T300/201 (Ti-B)					
568 C, 24 MPa, 25 min	212	36	3280	19	8
T Pitch/201 (Ti-B)					
568 C, 24 MPa, 25 min	313	34	1568	61	8
HM3000/201 (Ti-B)					
568 C, 24 MPa, 25 min	1014	40	369	98	8
HM3000/6061 (Ti-B)					
596 C, 24 MPa, 25 min	692	34	203		8
HM3000/1100 (Ti-B)					
607 C, 24 MPa, 30 min	621	30	176		8

TABLE 7

TRANSVERSE STRENGTHS OF CARBON/ALUMINIUM COMPOSITES PRODUCED BY VARIOUS METHODS

Composite	Production Method				% Volume Fibres %	Transverse Strength MPa	Reference
	Precursor	Consolidation					
		°C	MPa	Min.			
T50/201	Ti-B/LMI	570	24	25	32	24.4	8
T50/201	H pre-cleaning prior to Ti-B/LMI	574	24	25	35	21.0	8
T50/201	Pyrolytic carbon prior to Ti-B/LMI	568	24	25	36	19	8
T50/201	Sodium process	560	24	25	14	43.2	8
HM1000/201	Nickel plate/LMI	568	24	25	12	27.8	8
T300/201	Ti-B/LMI	568	24	25	36	55.8	8
T300/201	Pyrolytic carbon prior to Ti-B/LMI	568	24	25	44	10.3	8
HM3000/201	Ti-B/LMI	568	24	25	40	11.1	8
HM3000/6061	Ti-B/LMI	596	24	25	34	10.7	8
HM3000/1100	Ti-B/LMI	607	24	30	30	9.7	8
Thornel Type-P/201	Ti-B/LMI	568	24	25	34	28.9	8
T300/201	Pyrolytic carbon prior to Ti-B/LMI	—	—	—	—	10	6
T50/201	Ti-B/LMI	—	—	—	—	20	6
T50/201	Ti-B/LMI	—	—	—	—	34.5	26
T50/6061	Ti-B/LMI	575	21	40	27	15-37	30

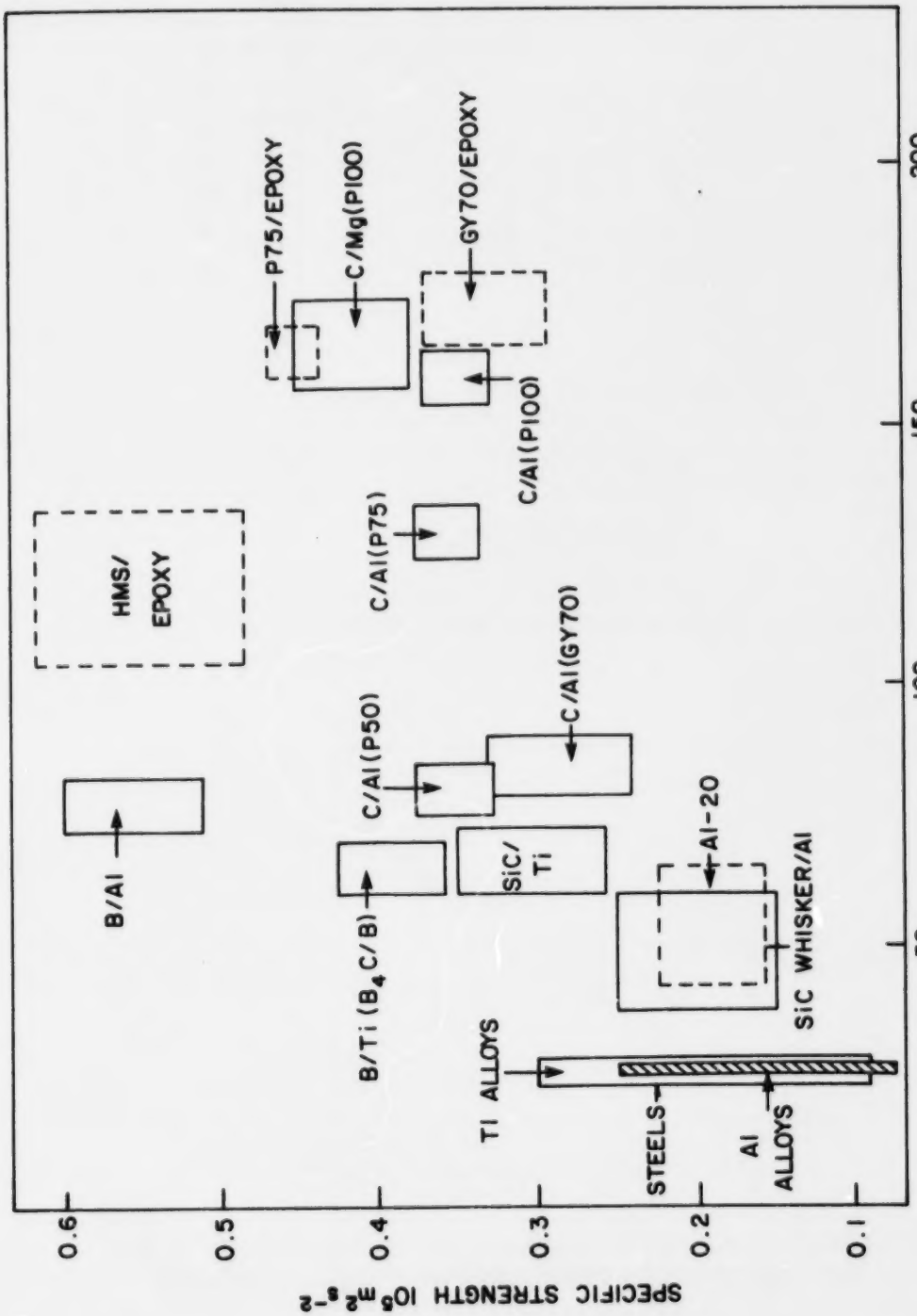

FIG. 1: SPECIFIC STRENGTH AND MODULUS OF SOME METAL MATRIX COMPOSITES COMPARED TO SOME OTHER MATERIALS[1]

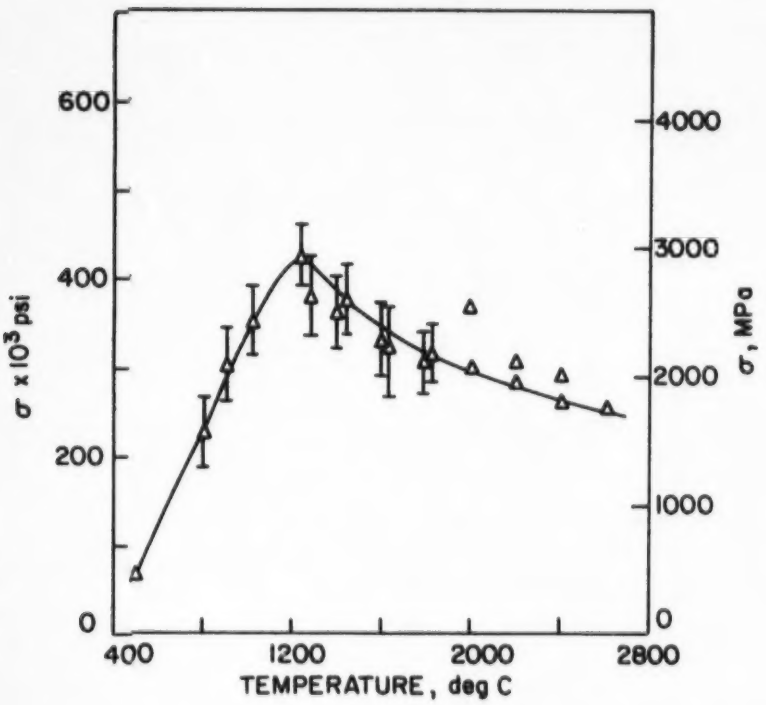

FIG. 2: ULTIMATE TENSILE STRENGTH OF PAN BASE CARBON FIBRES VERSUS HEAT TREATMENT (GRAPHITIZATION) TEMPERATURE[2]

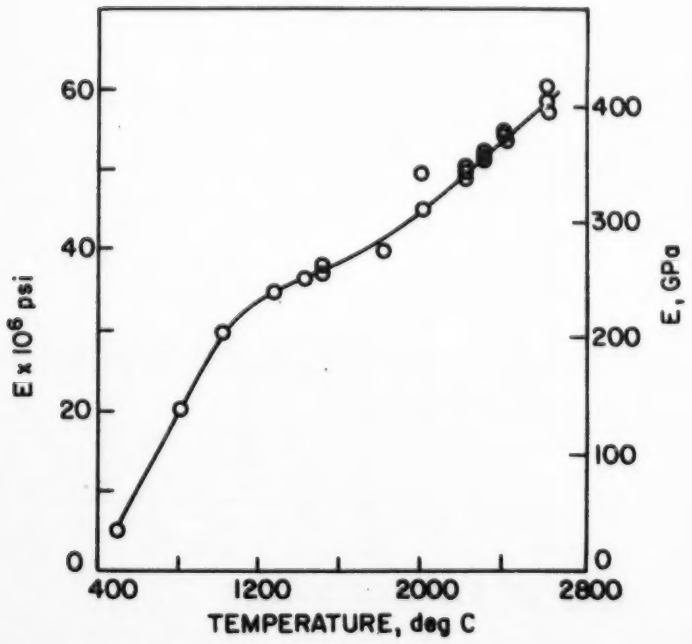

FIG. 3: TENSILE MODULUS OF ELASTICITY VERSUS FINAL HEAT TREATMENT (GRAPHITIZATION) TEMPERATURE FOR PAN BASE CARBON FIBRES[2]

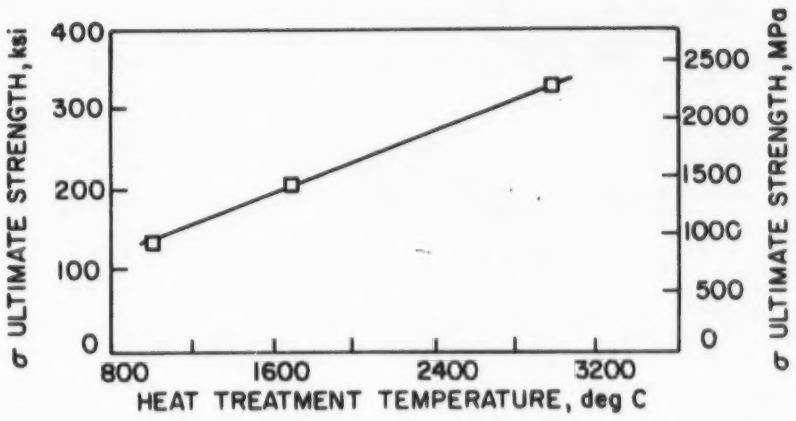

FIG. 4: ULTIMATE TENSILE STRENGTH VERUS FINAL HEAT TREATMENT (GRAPHITIZATION) TEMPERATURE FOR PITCH BASE CARBON FIBRES[2]

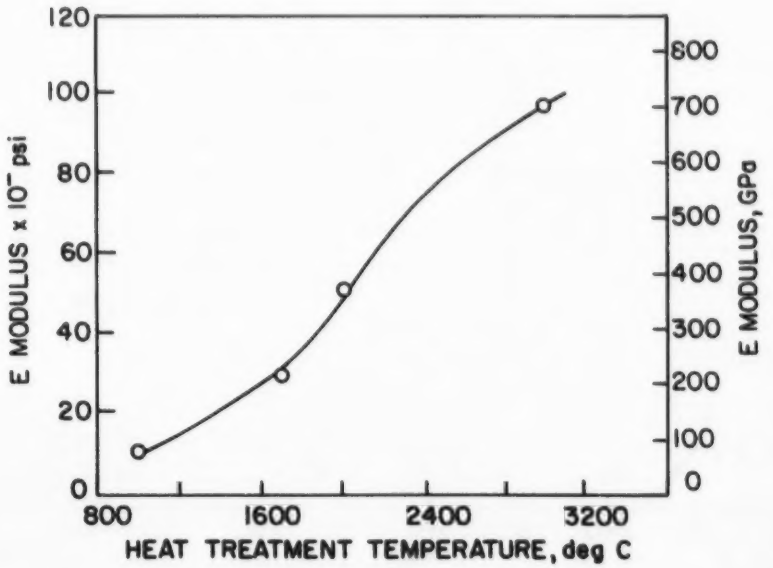

FIG. 5: TENSILE MODULUS OF ELASTICITY VERSUS FINAL HEAT TREATMENT (GRAPHITIZATION) TEMPERATURE FOR PITCH BASE CARBON FIBRES[2]

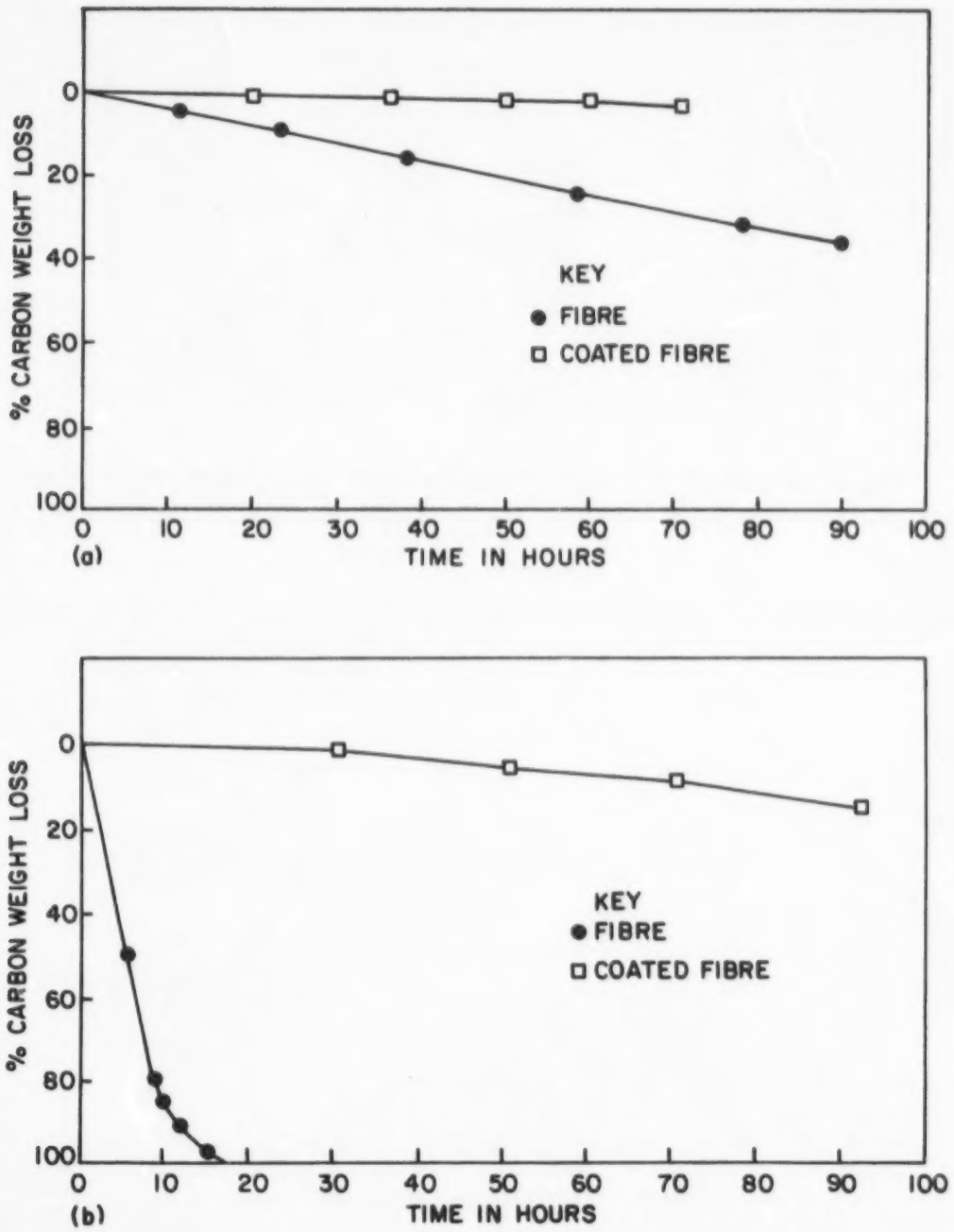

FIG. 6: PLOT OF TIME VERSUS TEMPERATURE FOR ALUMINIUM COATED AND UNCOATED HIGH MODULUS CARBON FIBRES EXPOSED IN AIR AT (a) 475°C AND (b) 525°C

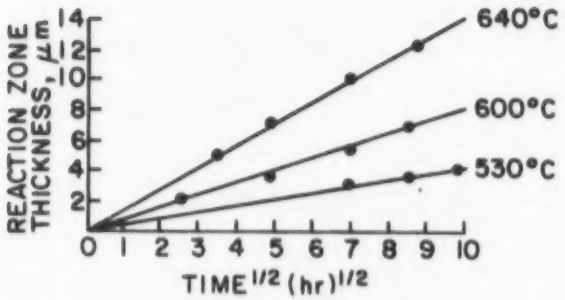

FIG. 7: THICKNESS OF THE REACTION ZONE AS A FUNCTION OF TEMPERATURE AND TIME FOR CARBON/ALUMINIUM SYSTEMS[30]

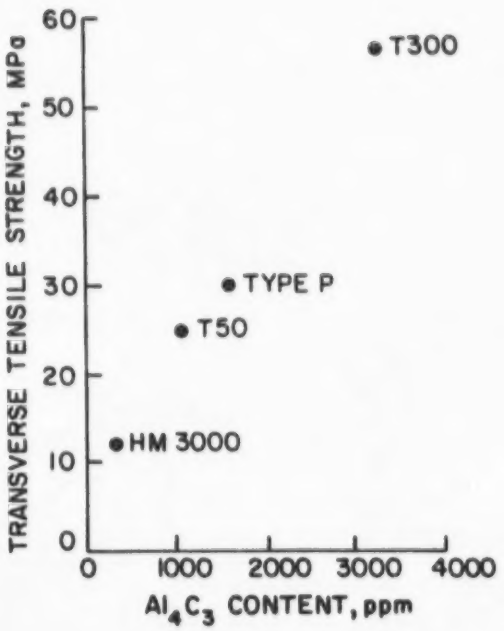

FIG. 8: RELATIONSHIP BETWEEN CARBIDE FORMATION AND TRANSVERSE STRENGTH OF VARIOUS CARBON FIBRE/201 COMPOSITES[11]

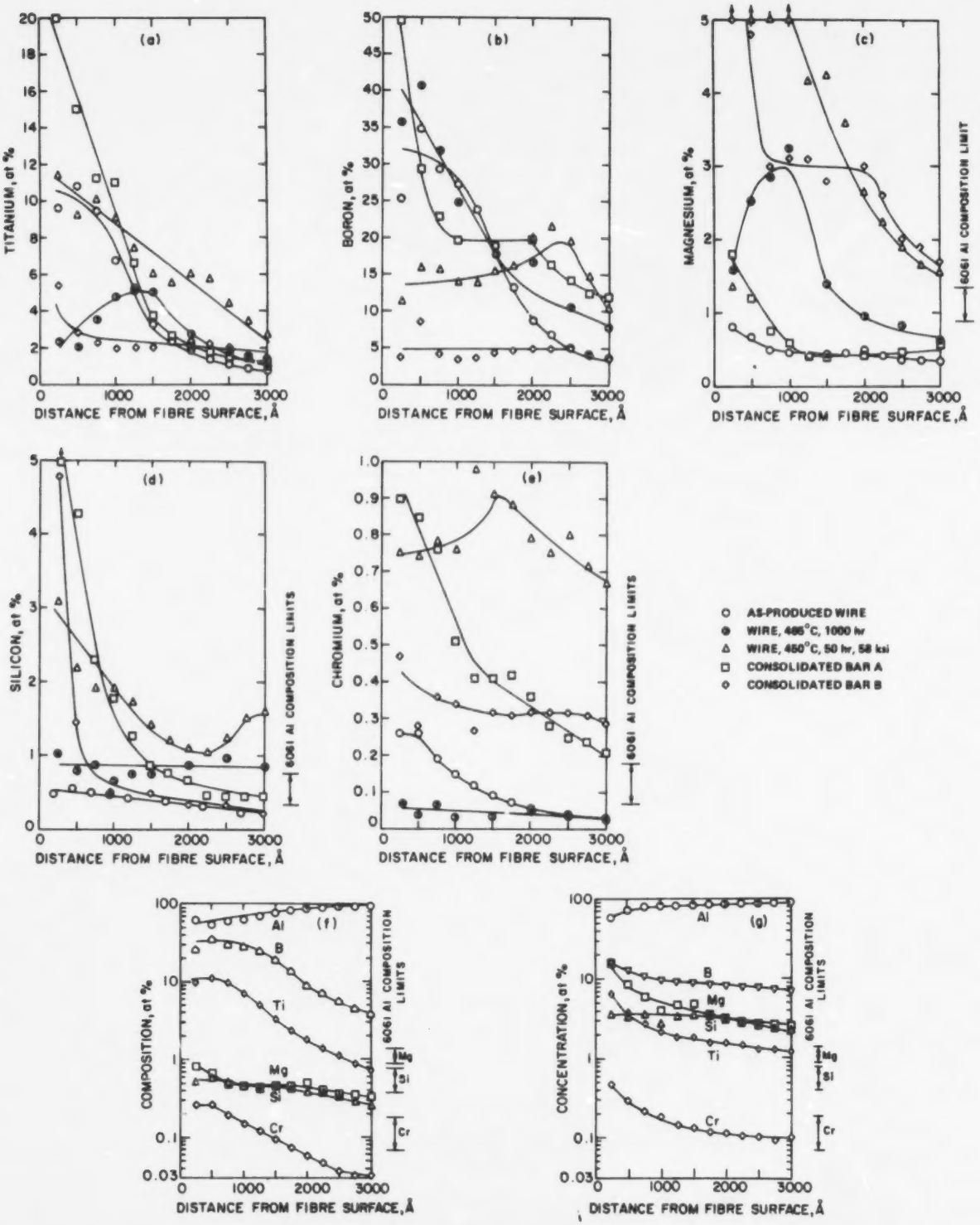

FIG. 9: IMMA ELEMENT PROFILES NEAR FIBRE SURFACE FOR T50/6061 COMPOSITE AFTER VARIOUS THERMAL AND MECHANICAL TREATMENTS. (a) TITANIUM, (b) BORON, (c) MAGNESIUM, (d) SILICON, (e) CHROMIUM, (f) AS RECEIVED PRECURSOR WIRE, (g) COMPOSITE CONSOLIDATED AT 600°C(35)

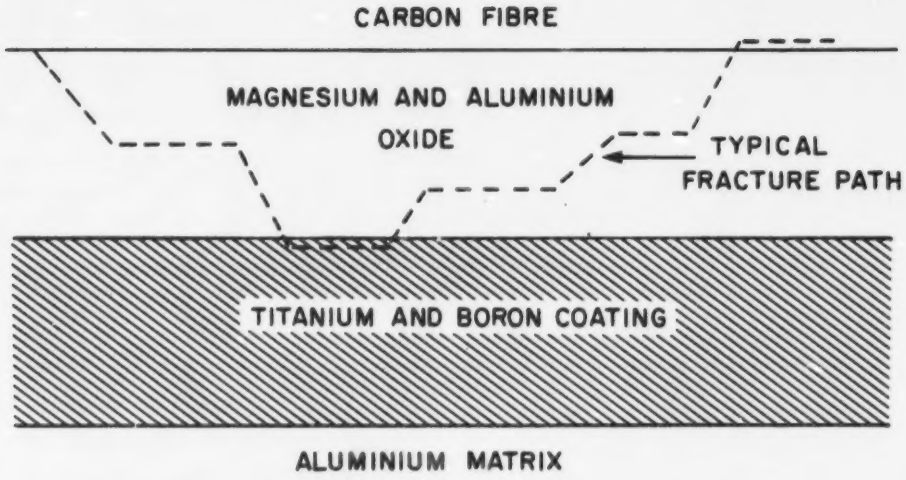

FIG. 10: FRACTURE PATH IN T50/201 COMPOSITE[6]

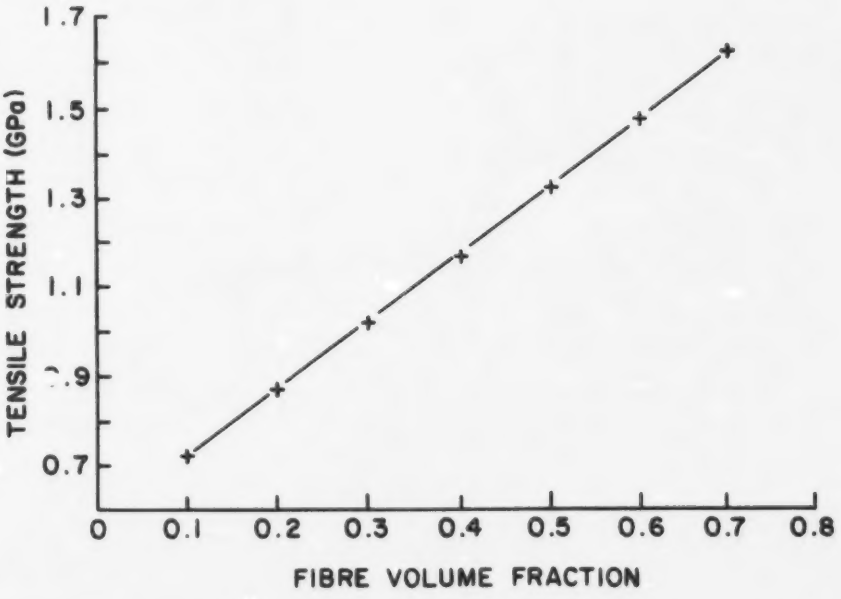

FIG. 11: PLOT SHOWING TENSILE STRENGTH AS PREDICTED BY THE RULE OF MIXTURES VERSUS VOLUME FRACTION OF FIBRES FOR P-100/7075 COMPOSITES

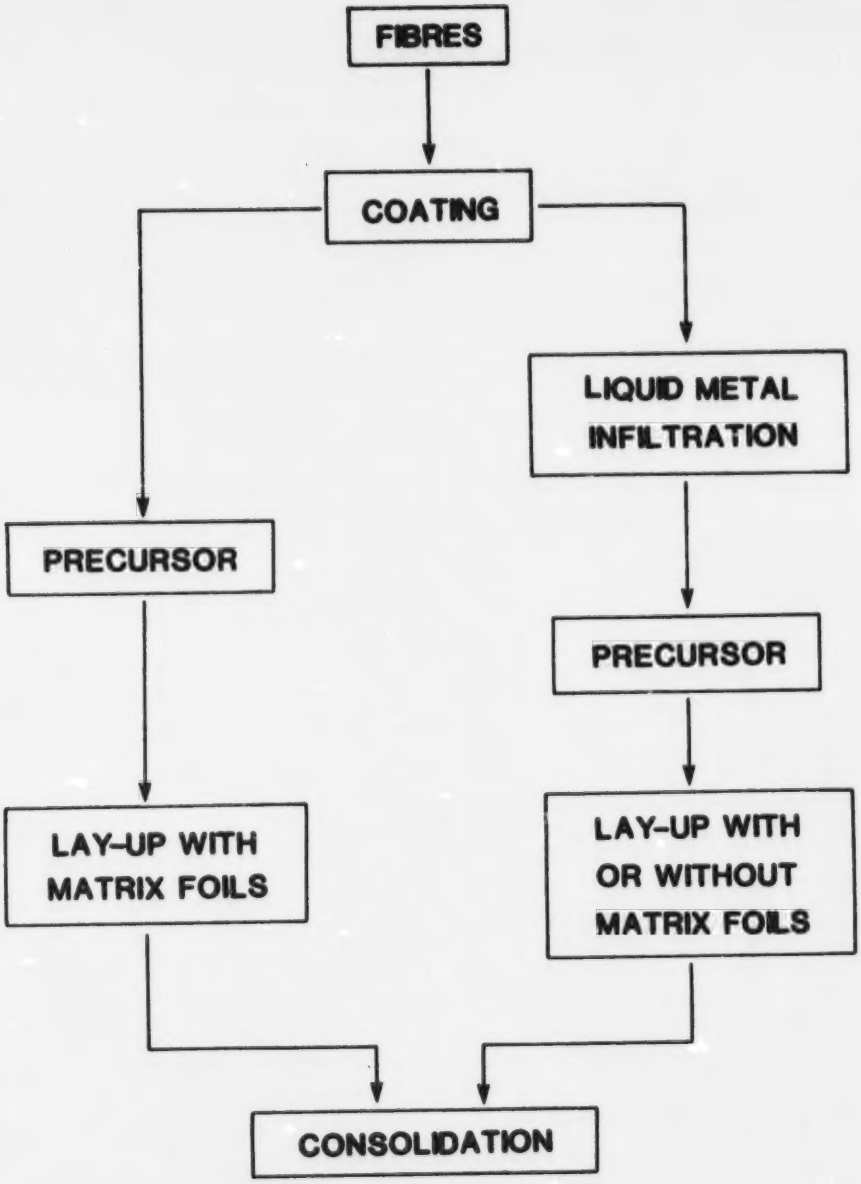

FIG. 12: SIMPLIFIED FLOW CHART OF THE PROCESSES INVOLVED IN THE PRODUCTION OF CARBON/ALUMINIUM COMPOSITES

REPORT DOCUMENTATION PAGE/PAGE DE DOCUMENTATION DE RAPPORT

REPORT/RAPPORT DM-4 1a	REPORT/RAPPORT NRC No. 23498 1b
REPORT SECURITY CLASSIFICATION CLASSIFICATION DE SÉCURITÉ DE RAPPORT Unclassified 2	DISTRIBUTION/DIFFUSION Unlimited 3

TITLE/SUBTITLE/TITRE/SOUS-TITRE (CLASSIFICATION)
Carbon Fibre Reinforced Aluminium Matrix Composites
A Critical Review
4

AUTHOR(S)/AUTEUR(S)
M.U. Islam, W. Wallace
5

SERIES/SÉRIE
Division of Mechanical Engineering Report
6

CORPORATE AUTHOR/PERFORMING AGENCY/AUTEUR D'ENTREPRISE/AGENCE D'EXÉCUTION
National Research Council Canada — Manufacturing Technology Centre
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DESCRIPTORS(KEY WORDS)/MOTS-CLÉS
1. Reinforced aluminium
2. Reinforced materials
3. Composite materials
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SUMMARY/SOMMAIRE
In this report, the state-of-the-art technology on the carbon fibre reinforced aluminium matrix composites is reviewed. The available raw material, various fabrication techniques along with their effects on the properties of the composites produced and the manufacturing problems encountered are described in detail. General trends are outlined and further research and development work on composites containing pitch base carbon fibres in a matrix of high strength aluminium alloy is recommended.
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